X-ray diffraction study of CdSe/BeTe nanostructures grown by MBE with stressor-controlled interfaces

R. N. Kyutt†⋆, T. V. Shubina‡, S. V. Sorokin‡⋆, S. V. Ivanov‡⋆, P. S. Kop'ev‡,
A. Waag‡, G. Landwehr⋆ and M. Willander⋆⋆

† Ioffe Physico-Technical Institute, St Petersburg, Russia
‡ Abteilung Halbleiterphysik, Universität Ulm, 89081 Ulm, Germany
⋆Physikalisches Institut der Universität Würzburg, D-97074 Würzburg, Germany
⋆⋆Physical Electronics and Photonics, Dept of Microelectronics and Nanoscience,
Chalmers University of Technology and Göteborg University,
S-412 96 Göteborg, Sweden

Self-formation of quantum-confined nanostructures of wide gap II-VI compounds (e.g. CdSe/ZnSe) is characterised by enhanced broadening (up to 10 monolayers (ML)) of the deposited strained insertions even when they are in a sub-monolayer thickness range [1, 2]. This results in smoothing a carrier localisation potential, preventing fabrication of real quantum dots (QDs) optically active up to the room temperature. Therefore, looking for a novel approach to the 0D-nanostructure fabrication, differing from the conventional Stranski–Krastanov growth mode, is of great importance. Recently, a stimulating effect of an intentionally-introduced super-strained fractional monolayer (FM) of BeSe — "stressor" — on the self-formation of CdSe QDs in ZnSe has been observed [3]. The stressor-assisted growth appears to be most effective for the Be-containing compounds with strong chemical bonds resulting in sharp nanostructure interfaces [4]. Novel CdSe/BeTe nanostructures grown using CdTe FM as a stressor [5] have demonstrated a possibility of CdSe nano-island ($<$ 10 nm) ordering and an interface-induced linear optical anisotropy due to using two binary compounds with no-common atoms [6].

In the paper we report on x-ray diffraction (XRD) study of the CdSe/BeTe nanostructures, intended to elucidate chemical composition of interfaces, playing a role of natural stressors, and process of their formation during molecular beam epitaxial (MBE) growth.

The study has been performed on CdSe/BeTe superlattice (SL) structures grown on GaAs(100) substrates at 300°C in a Riber 2300 MBE setup. RHEED specular spot intensity (SSI) oscillations were monitored to control the interface formation with a high accuracy [5]. Three sets of the structures were investigated. Two sets involve CdSe/BeTe SLs with CdTe interfaces intentionally formed by CdTe FMs ($<$ 0.3 ML) deposited prior the growth of each CdSe and BeTe layers. For the first set of the SL structures, CdTe growth at the first BeTe-CdSe interface was interrupted at a local RHEED SSI maximum and for the second SL set – at the SSI minimum [5]. In both sets the second interfaces are grown at the SSI maximum. SL samples of the third set contain only the CdTe FM insertions in BeTe, deposited till either the SSI minimum or the SSI maximum for the reference. The intended SL period is between 5–6 nm, a nominal thickness of the CdSe insertion is varied in the 0.4$-$1.5 ML range for different samples. The parameters of the samples are listed in Table 1. XRD measurement were performed on a Philips X-Pert Diffractometer using the symmetrical Bragg reflections 002 and 004. The double crystal diffraction (DCD) rocking curves (RCs) and triple crystal scans were recorded using $CuK_{\alpha 1}$ radiation.

The XRD curves measured demonstrate periodical intensity distribution with SL-satellites and well-resolved thickness fringes for most of the samples. We present here only 002 reflection RCs containing more details and having the more distinct satellites and fringes as compared to 004 ones, which is due to the larger value of the BeTe structural factor than those of CdTe and CdSe.

The accurate theoretical simulation of the RCs of CdTe/BeTe SLs (the samples of the third set) (see Fig. 1(a)) allowed us to determine the CdTe nominal thicknesses corresponding to both the

2

Table 1. Parameters of the samples used in the study.

Sample No	CdSe width (ML)	CdTe interface deposition	SL period thickness (Å)	SL(0) position (arc. sec)	ΔΘ deviation (arc. sec)	BeSe (ML)
1480	—	min SSI (0.08 ML)	39.3	433	—	—
1481	—	max SSI (0.2 ML)	44.9	205	—	—
1484	0.4	Max–max	45.9	147	229	0.45
1487	0.7	Max–max	50.8	25	229	0.48
1483	1.0	Max–max	56	310	716	1.12
1489	1.5	Max–max	57–66	−173	383	0.77
1502	0.7	Min–max	51	51	641	1.04
1505	1.0	Min–max	29.4	1610	2500	0.45
1503	1.5	Min–max	54	54	973	1.59

SSI maximum and minimum. The values were further employed in processing XRD data of other samples. One should stress that 004 RCs of the samples of the both first and second sets have a similar shape, while the 002 curves differ considerably by the satellite intensity. Samples of the first set exhibit a more strong SL(+1) satellite peak, whereas the SL(−1) one almost disappears (Fig. 1(b)). For samples of the second set RCs have a more symmetrical character (Fig. 1(c)). The values of the SL-period determined from simulation these RCs are also given in Table 1.

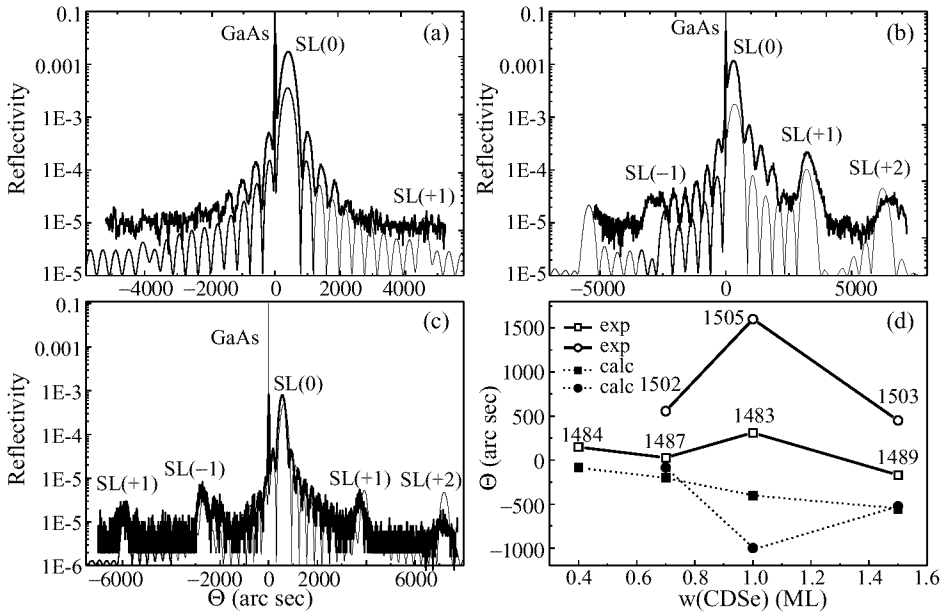Comparison of the experimental SL(0) peak positions with those expected from the measured

Fig. 1. Experimental (thick line) and simulated (thin line) DCD (002) rocking curves for (a) CdTe/BeTe SL (#1480); (b) CdSe/BeTe SL with the max-max CdTe interfaces (#1483); (c) CdSe/BeTe SL with the min-max CdTe interfaces (#1502). Angular position of the SL(0) peak at 002-rocking curves for the samples of the first set (squares) and the second set (circles) as a function of CdSe nominal thickness is presented in (d).

SL period and intended composition (CdSe and CdTe contents) shows that the experimental values are strongly deviated to the positive angles. The dependence of the angular deviation on the CdSe-content (Fig. 1(d), Table 1) has a non-monotonic character: an abrupt shift of the SL(0) peak position for samples #1483 and #1505 with $w \sim 1.0$ ML is observed as compared to others in each set following a monotonically decreasing dependence. The deviation is obviously larger for the second set samples, reaching for #1505 the value which gives the SL(0) position at the angle higher than that corresponding to a pure BeTe layer.

These facts make us suppose that there exists an additional component with the lattice parameter smaller than that of BeTe in the SLs studied, which can be BeSe only. Be-Se interface bonds can arise at the BeTe surface free from the intentionally deposited CdTe islands due to exchange reactions between Se atoms impinging onto the surface with the BeTe lattice. These reactions are strongly shifted to the BeSe formation owing to its much higher binding energy. Obviously, the SLs of the second set with the smaller CdTe coverage (SSI minimum) of BeTe surface should display higher contribution of BeSe interface bonds. To explain the maximum of BeSe-bond contribution at $w(CdSe) \sim 1$ ML one should take into account that CdSe is deposited in MBE mode under strong Se-rich conditions. The excessive Se atoms may interact with the free BeTe surface, resulting in the BeSe bonds formation, faster than a CdSe deposition rate. However, this process comes to saturation soon, while CdSe amount continuously increases.

Simulation including BeSe contribution provides more satisfactory coincidence of the experimental and calculated RCs (Fig. 1(b) and (c)), although does not reproduce all the details of the experimental curves, particularly the shape of the satellite peaks. Probably, there are some additional peculiarities of the interface structure, like e.g. lateral composition distribution, which influence the diffraction intensity distribution.

In conclusion, we perform a detail XRD study of the novel type-II CdSe/BeTe nanostructures formed by stressor-assisted self-formation. High sensitivity of the diffraction curves to interface types is revealed. The model developed for a simulation of rocking curves includes spontaneous formation of BeSe-enriched interface formation when the smaller amount of CdTe FM stressor as compared to that corresponding to a maximum RHEED SSI is intentionally deposited.

Acknowledgements

References

[1] R. N. Kyutt, A. A. Toropov, S. V. Sorokin, T. V. Shubina, S. V. Ivanov, M. Karlsteen and M. Willander, *Appl. Phys. Lett.* **75**, 373 (1999).

[2] N. Peranio, A. Rosenauer, D. Gerthsen, S. V. Sorokin, I. V. Sedova and S. V. Ivanov, *Phys. Rev. B* **61 (23)**, 16015 (2000).

[3] M. Keim, M. Korn, F. Bensing, A. Waag, G. Landwehr, L. W. Molenkamp, S. V. Ivanov, S. V. Sorokin, A. A Toropov, T. V. Shubina, P. S. Kop'ev, J. Seufert, R. Weigand, G. Bacher and A. Forchel, *J. Cryst. Growth*, **214/215**, 684 (2000).

[4] Th. Walter, A. Rosenauer, D. Gerthsen, F. Fischer, R. Gall, Th. Litz, A. Waag and G. Landwehr, *Microscopy of Semiconducting Materials*, UK: Institute of Physics Publishing, Oxford, 1997, p. 315.

[5] S. V. Ivanov, G. Reuscher, T. Gruber, T. Muck, V. Wagner, J. Geurts, A. Waag, G. Landwehr, T. V. Shubina, N. A. Sadchikov, A. A. Toropov and P. S. Kop'ev, *Proc. 8th Int. Symposium "Nanostructures: Physics and Technology"*, St Peterburg, 2000, p. 98.

[6] T. V. Shubina, A. A. Toropov, S. V. Sorokin, A. V. Lebedev, S. V. Ivanov, A. Waag, G. Landwehr and L. W. Molenkamp, *Abstracts of 25th Int. Conf. on the Physics of Semiconductors*, Ocaka, Japan, 2000, p. 219.